

ABSTRACT

Title of dissertation:

CHARACTERIZATION OF MECHANICAL PROPERTIES AND DEFECTS OF SOLID-OXIDE FUEL CELL MATERIALS

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Solid-oxide fuel cells (SOFCs) have the potential to help meet global energy demands by efficiently converting fuel to electricity. The technology currently requires high temperatures and has reliability limitations. A critical concern is the structural integrity of the cell after redox cycling at operating temperatures. As new materials are developed reduce operating temperatures and improve redox stability, the effect of the environment on their mechanical properties must be studied. Ceria-based systems have allowed the operating temperature to be decreased to the 600 °C range. For this reason, a three-point bend apparatus was developed which could test materials at 650 °C in reducing environments.

Using this apparatus, it was shown how pore shape effected strength of porous gadolinium doped ceria (GDC) at 650 °C. The strength of Ni-GDC half-cell coupons showed no dependence on loading orientation at elevated temperatures in air but were 47% weaker when the electrolyte was placed in tension under H₂. It was also determined that a reduced Ni-GDC cell could be exposed to air for an extended

period of time and reheated under H₂ with no effect to the strength.

A new anode material, SrFe_{0.2}Co_{0.4}Mo_{0.4}O_{3-δ} (SFCM), was investigated for oxygen non-stoichiometry. SFCM maintains phase purity under reducing conditions, with little changes to lattice parameter between oxidation and reduction. Under oxidation SFCM forms phase impurities. SFCM supports the formation of a large number of oxygen vacancies with a $\Delta\delta = 0.176$ due to the low enthalpy of formation for oxygen vacancies at 39.1 kJ mol⁻¹.

Fracture toughness of SFCM was determined to be (0.124 ± 0.023) MPa $\sqrt{\text{m}}$ in air at room temperature and (0.286 ± 0.038) MPa $\sqrt{\text{m}}$ at 600 °C. The strength of SFCM-GDC half-cells increased by 31% upon heating to 600 °C after which reduction decreased strength by 29%. Reduction and redox cycling were shown to only decrease the characteristic strength, not alter the structural flaw distribution.

This work...

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by

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Dedication

I dedicate this to my wife, Bethany, who has encouraged and supported me throughout the process, and to my two wonderful daughters, Lucy and Madeleine. You brighten my life in ways I cannot begin to explain.

Acknowledgments

I would like to start off by thanking my graduate advisor, Dr. Eric Wachsman, for his support during my graduate career. He allowed me the freedom to direct the details of my research and the support to help me reach my goals. He was also supportive of me when my family began to grow through marriage and children, allowing me to continue to develop my personal life while pursuing my degree.

More thanks than I can express is due to my wonderful wife, Bethany and now our daughters, Lucy and Madeleine. When Bethany and I first met, I was still deciding between gradate programs, and now, after many life changes, she continues to challenge and support me in my endeavors. Lucy and Madeleine have provided me with impetus to continue pushing through the challenges and to strike a balance between my personal and professional pursuits.

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I have been able to interact with, I would like to especially thank Dr. Isabel Lloyd who has graciously given me her advice, help make introductions, and passed along opportunities to me when they come her way.

Somehow a small group of my friends of undergraduate have accumulated in the area. It has been a blessing to have them near by, always willing to help out when need be or to just spend time with each other. Dave and Katie Shahin are a large part of what brought me to the University and it has been a blast to go through the program with Dave. Andrew Dunkman has helped me in so many ways, from apartment hunting a thousand miles away to programming questions, for which I express my thanks. And finally, Angela Rudolph has been very supportive of me and my family getting through graduate school.

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List of Abbreviations and Symbols

p_{O_2} partial pressure of oxygen

GDC gadolinium doped ceria

LSM lanthanum strontium manganite

MIEC mixed ionic-electronic conductor

SFCM $SrFe_{0.2}Co_{0.4}Mo_{0.4}O_{3-\delta}$

SOFC solid-oxide fuel cell

TPB triple phase boundary

YSZ yttria-stabilized zirconia

Chapter 1: Introduction

1.1 Motivation for Solid-Oxide Fuel Cell Research

The world depends on fossil fuels for its daily energy needs. This is a fact that is going to stay with us for the foreseeable future. By 2040, it is estimated that fossil fuels will still supply 78% of total energy demand. [1] Traditionally, the annual increase in demand is met by an increase in supply of fuel. This increase is supplied by developments in mining and drilling operations, although recent advances have been found to be controversial, such as hydraulic fracturing. [2, 3] These concerns do not even account for the increased emissions caused from combustion of fossil fuels and their effect on the global environment. [4–6]

An alternative to increasing the supply and use of fossil fuels is to increase the efficiency of converting the fuel into useful work. Combustion generators convert chemical energy to thermal energy, then to mechanical energy, and finally into electrical energy. Each energy conversion step has intrinsic losses which limit efficiency. Solid-oxide fuel cells (SOFCs) have the ability to solve this for applications where electrical power is desired. SOFCs allow for side stepping of these losses by allowing for the direct conversion of chemical energy to electrical energy. Power generation plants have efficiencies around 30%, while a stand-alone SOFC generator can convert

fuel to electricity at 45 to 65% efficiency. [7,8] SOFCs are able to run on a variety of fuel sources, such as hydrogen, methane or even biogas. [9] This fuel flexibility and higher efficiency positions SOFCs to bridge the gap in the energy economy as it transitions from fossil fuels to renewable sources. However, many technical hurdles still remain which currently limit SOFCs' viability and must be addressed.

1.2 Solid-Oxide Fuel Cell Operation

The layout for a hydrogen-fueled SOFC is presented in Figure 1.1. Fuel is exposed to the anode side of the cell, while an oxygen rich gas (usually air) is exposed to the cathode side. Separating the two sides is the cell itself, with anode, electrolyte, and cathode. The electrolyte is only conductive to oxygen ions, not electrons. A perfect electrolyte would have no electronic conduction, allowing no leakage current from anode to cathode and forcing all electrons to travel via an alternative route. At the cathode, diatomic oxygen gas will disassociate into oxygen ions, each with a 2– charge from the addition of electrons from the cathode to the oxygen. The oxygen ions then travel though the electrolyte and react with the fuel at the anode, freeing the previously captured electrons. To maintain charge balance, electrons must travel in the opposite direction to the cathode. Because the electrolyte only allows oxygen ions to diffuse though it, the electrons travel though an external circuit performing useful work in the process. A limiting factor in this movement is the ability of oxygen to transport into, through, and out of the cell. As the rate of oxygen transport is increased, the ability for electrons to perform work also increases.

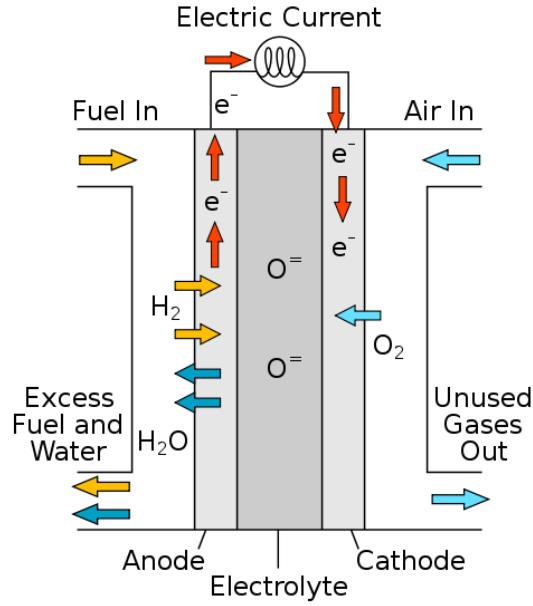


Figure 1.1: Diagram showing flow of materials in the operation of an SOFC. [10]

The driving force for the movement of oxygen ions and electrons comes from the chemical potential gradient across the cell. When the anode is exposed to fuel, it creates a low partial pressure of oxygen (p_{O_2}) in comparison to the cathode, which is exposed to air. This gradient drives the motion of oxygen through the cell from the high p_{O_2} of the cathode to the low p_{O_2} of the anode, and generates electric current in the opposite direction to maintain a charge balance. The electrical potential at open circuit conditions (voltage with no current draw) is expressed by the Nernst equation, given in Equation 1.2 for the reaction given in Equation 1.1, where f_A is the fugacity of species A, E^o is the standard potential for the reaction, and n is the number of charges involved in the reaction, R is the ideal gas constant and F is Faraday's constant. [11] Commonly, the fugacity is replaced by the partial pressure of

the species. For the case of a hydrogen-fueled SOFC, Equation 1.2 can be simplified to Equation 1.3. [12] From these equations it can be seen that the partial pressures of the involved species have a large role to play in the open circuit voltage of the cell. It is critical that the fuel and air remain separated by the cell. If there is a leak either through the cell or around it, then the electric potential across the cell will decrease harming performance.



$$E = E^o - \frac{RT}{nF} \ln \left(\frac{f_C^c f_D^d}{f_A^a f_B^b} \right) \quad (1.2)$$

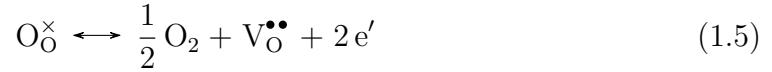
$$E = E^o + \frac{RT}{2F} \ln \left(\frac{p_{H_2}(p_{O_2})^{1/2}}{p_{H_2O}} \right) \quad (1.3)$$

While the electric potential is established by the partial pressure gradient, the electric current is limited by the ionic conduction of oxygen ions from the cathode to the anode. In ionic conduction, oxygen ions hop from site to site via oxygen vacancies, locations where an oxygen atom is missing from the site it usually occupies in the crystal lattice. Because conduction depends on vacancies, the more vacancies that are present in the lattice, the more available sites for oxygen to move between, the greater the conductance, and the greater power output of an cell. The relationship between oxygen vacancy concentration and oxygen conductance is given by Equation 1.4, where σ_i is the ionic conductivity, $[V_O^{••}]$ is the concentration of oxygen vacancies in the lattice, μ_O is the ionic mobility of the oxygen, and Z_e is the charge of the conducting species. [13] Thus, by adding more oxygen vacancies the ionic conductivity increases. While additional oxygen vacancies are a benefit to the electrical performance of the

cell, it can have detrimental effects to the mechanical properties.

$$\sigma_i = [V_O^{\bullet\bullet}] Z e \mu_O \quad (1.4)$$

During operation, oxygen vacancies are created by removal of lattice oxygen from the anode as it reacts with the fuel, as shown in the half-cell reaction of Equation 1.5. This reaction takes place on the surface of the anode, but for convenience it is written where oxygen has associated back into a diatomic state and the fuel subsequently react with it. This reaction of fuel with oxygen creates an environment with a very low p_{O_2} at the surface of the anode. A steady state concentration of oxygen vacancies will be reached as the cell comes into equilibrium with the environment, but this concentration will depend on the exact environmental, material, and performance conditions of the cell.



The cathode and anode materials must facilitate the incorporation of oxygen into the cell and the reaction with the fuel. They must also serve to transport the electrons into and out of the cell as oxygen changes state. As a result, the anode and cathode must possess both electrical and ionic conductivity. To achieve this, a mixture of the electrolyte material and an electronically conductive material are used to create these structures. This mixture of materials creates specific sites where the cell is active, known as triple phase boundarys (TPBs) and demonstrated in Figure 1.2. The TPB is the location where gas, electronic conductor, and ionic conductor meet and oxygen can incorporate into the cell or react with the fuel. To assist gas diffusion to the TPB and to maximize the length of TPBs, pores are added, greatly

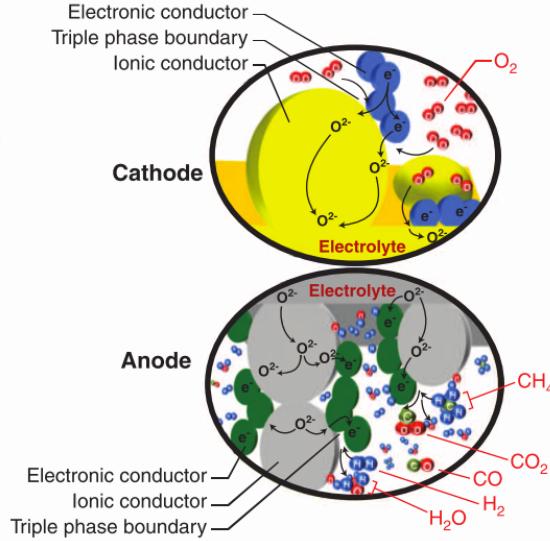


Figure 1.2: Diagram showing the triple phase boundary and the importance as being the site where incorporation and reactions occur. [7]

increasing the available surface area of the electrodes. The more TPBs that are present, the more exchange can occur between the gas and the cell at any given time and the current output of the cell is increased. This added porosity does not come without the disadvantage of decreasing the mechanical strength of the cell. [14, 15]

An operating SOFC would consist of a series of individual cells stacked together for combined power output. The cells would be sealed between current collectors and interconnects and heated to operating temperature. The elevated temperature enables oxygen conduction through the electrolyte by increasing vacancy concentration and mobility, but decreases the theoretical voltage. Historically, SOFCs operated near 1000 °C, but advances in materials now allow cells to operate at 650 °C and development continues to decrease the operating temperature. At lower operating temperatures it becomes easier to create better seals and the stack can be built from

more readily available materials. Additionally, the cells are subjected to less thermal expansion, reducing the stresses on the cell.

1.3 Solid-Oxide Fuel Cell Materials

Each structure of an SOFC is fabricated from combinations of materials designed to optimize that part's function then laminated with the other structures to create the cell. For example, a cell can be comprised of a nickel metal-gadolinium doped ceria (GDC) anode, a GDC electrolyte, and then a lanthanum strontium manganite (LSM) cathode. [16, 17] The individual parts are cast into thin flexible sheets (tapes) from bulk powders, with additives such as pore formers where appropriate, and then laminated together and co-fired to create a single cell. As a result, a completed cell has distinct layers in it where the material and properties abruptly change. This composite structure can then have inter-diffusion between layers, smoothing out the abrupt changes, but creating new structures and compositions that were not present upon lamination. [18]

The traditional material used for electrolytes in SOFCs has been yttria-stabilized zirconia (YSZ). YSZ is very stable, has a ratio of ionic conductivity to total conductivity (transference number) of approximately 1, and good conductivity at temperatures in excess of 800 °C and up to very low p_{O_2} s. A high transference number minimizes leakage current due to electronic conduction in the electrolyte, maintaining cell performance. As mentioned previously, GDC is another choice for an electrolyte. Other rare-earth doped cerias have been used as electrolytes, but

GDC tends to outperform them under a variety of conditions. At 600 °C, GDC exhibits an order of magnitude higher conductivity than YSZ, but has a smaller electrolytic domain where the transference number is near 1. [19] Under conditions of high temperature or very low p_{O_2} , GDC will start to leak electrons through the electrolyte, reducing cell performance. As SOFC operating temperatures are reduced for improved efficiency, GDC is favored for the electrolyte material.

The anode usually consists of a mixture of the electrolyte material, which provides good ionic conductivity, with an electronic conductor that can facilitate fuel oxidation reactions on the surface. In combination with an ionic conductor it creates a cermet which functions as the anode. Nickel has commonly been used in the anode due to its high conductivity and catalytic properties. A challenge with nickel is that at ambient conditions it oxidizes into nickel oxide, but at operating conditions it reduces to the desired nickel metal. There is a large lattice parameter change between nickel and nickel oxide, such that, upon reduction it decreases volume by over 15%. As a result, the overall porosity of the anode layer increases as the SOFC is put into service and the anode is exposed to reducing conditions. [20, 21]

This increased porosity can greatly decrease the flexural strength and modulus of the anode and of the overall cell. [22, 23] Additionally, nickel is prone to poisoning from sulfur contaminants in the fuel, so alternative anodes are of interest for use in SOFCs.

As an alternative, other metal-ceramic systems have been researched to serve the function of the anode. These have included copper, cobalt, and platinum systems, but each suffer from a significant drawback such as coking, long-term performance

degradation or cost. A different approach is to use an all-ceramic anode, made of a mixed ionic-electronic conductor (MIEC). Perovskite materials have easily been developed into MIECs with the use of a transition metal occupying the octahedral B-site. $\text{SrFe}_{0.2}\text{Co}_{0.4}\text{Mo}_{0.4}\text{O}_{3-\delta}$ (SFCM) is a recently developed double perovskite MIEC with multiple dopants on the B-site with a high conductivity of 30 S/cm. [24–26] Due to the all-ceramic nature of these anodes, their thermal expansion coefficients better match that of the other ceramic components of the cell, reducing sintering stresses and improving redox cycling durability. Conversely, all-ceramic anodes tend to have lower conductivity and catalytic performance.

This work focuses on the structures in the cell which give mechanical support, the anode and electrolyte. Specifically, Ni-GDC and SFCM-GDC anodes with GDC electrolytes will be investigated.. While YSZ systems have been studied extensively, the mechanical properties of GDC systems have not been throughly explored. As SOFCs are developed which operate at lower temperatures, GDC is the preferred choice for an electrolyte material due to its high conductivity. Ni-GDC serves as a reliable starting place to characterize the mechanical properties of the anode, focusing on the changes which occur during reduction as NiO reduces to Ni metal. SFCM being a new material has no available data on the mechanical properties, let alone the non-stoichiometry which occurs in the material under reducing conditions.

Various cathode materials exist which assist in the incorporation of oxygen into the lattice. The cathode must have many similar properties to that of the anode, but be stable in oxidizing environments instead of reducing. As a result, perovskite structures again are found in many of the materials. LSM is one of the most

commonly used materials, but others include doped SrCoO_3 , $\text{La}_{0.6}\text{Sr}_{0.4}\text{Co}_{0.2}\text{Fe}_{0.8}\text{O}_3$, $\text{Ba}_x\text{Sr}_{1-x}\text{Ti}_{1-y}\text{Fe}_y\text{O}_{3-y/2+\delta}$. [27–30] The cathode does not tend to play a structural role in the fabrication of a cell, and thus is not investigated or developed in this work.

Outside the of the cell itself, several materials are needed to create a functioning stack. Metal interconnects support individual cells, holding them together, creating connections for the current to flow though current collectors and gas to flow though channels to the the cell. In intermediate and high temperature stacks, expensive metal alloys such as Inconel must be used to survive the extreme heat. As temperatures are lowered to below 600 °C more common metals can be used for interconnects, such as steel. Sealing materials are another critical component to assembly of an SOFC stack. Glasses with transition temperatures near the operating range are used at high temperatures as they can soften and flow into place creating gas tight seals. At lower temperatures new options are available, such as YSZ, vermiculite or mica gaskets.

1.4 Mechanical Properties

Ideally, an SOFC is as thin as possible to minimize diffusion path lengths and ionic resistances in the cell. [31] Realistically, the cell must be able to withstand the stresses of being manufactured, sealed, heating, and use. This means that a compromise must be made as to how the cell is supported and which components do the supporting. Traditionally, electrolyte supported cells were used with YSZ electrolytes where the thick dense electrolyte provides the structural support for the

cell. Recently anode supported cells made from GDC have been able to provide lower resistances due to the thinner electrolyte while adequately supporting the cell with a thicker anode which fuel can diffuse further into. [32, 33] Now the anode layer with its porosity, must support the majority of the stresses the cell is subjected to.

Ceramic materials used in SOFCs mechanically fail when stresses concentrate at a pre-existing microstructural flaw, overcoming the strength of the material, propagating a crack though the cell. Stresses can be intrinsic to the cell, left from lamination, sinter or cooling, or they can be applied externally during the sealing or operation of the cell. Microstructural flaws could be intentional with the added porosity or unintended flaws such as inclusions, contamination or grain agglomeration. These microstructural flaws are locations where stresses are concentrated and lead to the ultimate failure of the material.

Stresses are concentrated most by large, sharp featured flaws. As a result, in a uniform stress field, the largest flaw will usually cause failure. Ceramic materials will tend to fail where the sample is under tensile load and at a location near the point of maximum applied stress. Samples, even if processed together from the same raw materials, will each have a random sampling of flaws from the total batch, and a particular one which causes failure. No matter the care and attention paid to processing, some distribution of flaws will exist. The analysis of the distribution of flaws from a batch of samples is known as Weibull analysis and is explained in detail in Appendix ???. To perform Weibull analysis a large number of samples must be tested, but as a result the overall size/shape distribution of flaws can be obtained. This allows for the observation of the flaw distribution and if it changes at different

points or if another phenomenon is occurring.

Most of the materials used in SOFCs have crystal structures which promote the generation of oxygen vacancies. As oxygen vacancies are produced, the inter-atomic bonding of the structure changes, which can decrease the elastic modulus and fracture toughness of the crystal. [34, 35] If the strength of the atomic bonds weakens on average, due to added vacancies, it then follows that the elastic modulus and fracture toughness of the crystal would also decrease. This relationship has been shown to fit for single grains of GDC using nanoindentation, but this does not necessarily hold true for an actual cell. [36] Grain boundaries can play a large role in the mechanical properties of a bulk sample as they can impede or enhance the propagation of a crack. For this reason microstructure combined with environmental conditions can play a large role on the overall mechanical properties of a fuel cell.

The first external stress applied to the cell, and the one most likely to result in failure, is the sealing of an individual cell into a stack, creating a gas tight seal between the anode and cathode sides. In this process, multiple cells are sandwiched with gasket seals and interconnects between them and compressed until a gas tight seal is achieved. While it is a compressive force that is applied, the cells are never perfectly flat, resulting in a flexural stresses as the cell is pressed flat. Figure 1.3 highlights this fact by using an optical profilometer to measure the flatness of a 10 cm by 10 cm Ni-GDC/GDC half-cell. It can be seen that the cell curves by over a millimeter, mostly at edges, where the seal would be taking place. It is for this reason that the study of the failure of SOFC materials utilizes flexural testing.

To help develop SOFCs into a viable technology which can stretch the gap in

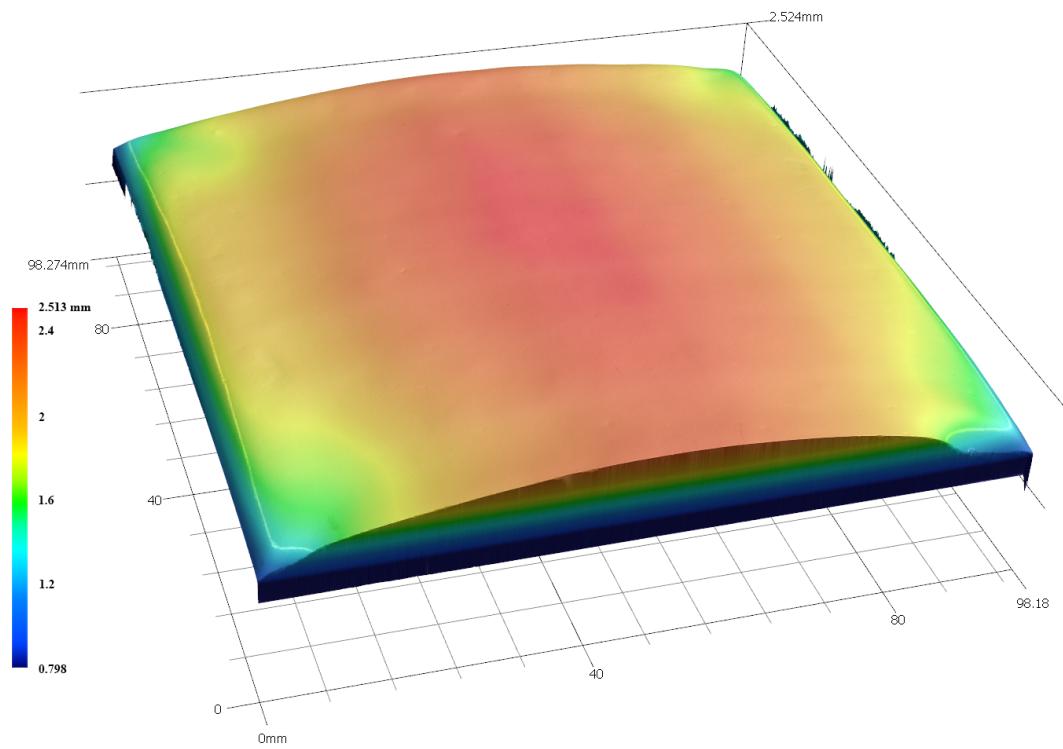


Figure 1.3: Optical profilometer measurements of a 10 cm by 10 cm Ni-GDC/GDC half-cell where the z-axis has been magnified 6X to highlight curvature

the energy economy, research is needed to improve the reliability of the cells during manufacturing and operation. This work aims to develop the understanding of fracture mechanics of real-world structures, by understanding how temperature and partial pressure of oxygen affect the strength, fracture toughness, and flaw distributions of SOFC components. This is done by analyzing the flexural properties of bulk bars and half-cell coupons at ambient conditions and at conditions similar to those found in an SOFC. Microscopy was used to look for changes in microstructure after various treatments in addition to Weibull analysis. Additionally, thermogravimetry, conductivity, and temperature programed desorption were used to understand the atomic defect structure of a new material to relate it to the mechanical properties.

Chapter 2: Experimental Procedures

2.1 Sample Preparation

2.1.1 $\text{SrFe}_{0.2}\text{Co}_{0.4}\text{Mo}_{0.4}\text{O}_{3-\delta}$ Powder

$\text{SrFe}_{0.2}\text{Co}_{0.4}\text{Mo}_{0.4}\text{O}_{3-\delta}$ (SFCM) was created from stoichiometric amounts of strontium carbonate (SrCO_3 , Sigma-Aldrich), iron oxide (Fe_2O_3 , Sigma-Aldrich), cobalt oxide (Co_2O_3 , Inframat Advanced Materials), and molybdenum oxide (MoO_3 , Alfa-Aesar) using conventional solid-state methods. The constituents were ball milled for 24 hours in ethanol and dried using a 100 °C oven. Afterwards the powder was calcined at 1100 °C for four hours.

2.1.2 Bars

GDC bars of varying porosity were fabricated using uniaxial pressing in a rectangular die. GDC10 ($\text{Ce}_{0.9}\text{Gd}_{0.1}\text{O}_2$) powder was mixed with the desired volume of poly(methyl methacrylate) (PMMA) microspheres (5 μm diameter) or graphite flake to create the green body, with 0.1wt% of polyvinyl butyral added to aid in pressing and a drop of fish oil to act as a dispersant during milling. The two different pore formers were used to create differing pore geometry. The PMMA or graphite

was removed by a 400 °C pre-sintering step, and the remaining ceramic structure was sintered at 1500 °C for 4 hours. Archimedes density measurements were used to confirm porosity percent. Samples were polished with 600 grit sandpaper on all sides before testing.

Dense SFCM bars were used to maximize the total mass and mass changes during thermogravimetric testing. Samples were made by combining SFCM powder with 0.6 wt% polyethylene glycol 600, 1.8 wt% ethylene glycol, and 0.6 wt% glycerol in isopropyl alcohol and ball milling overnight. After drying at 100 °C, the powder was ground by mortar and pestle, then pressed uniaxially into rectangular bars at 30 MPa, then isostatically pressed at 30 MPa. Bars were sintered by heating to 400 °C for one hour and then 1340 °C for four hours, using a 3 °C min⁻¹ heating and cooling rate. This produced bars with 97% theoretical density by Archimedes' technique.

To create fracture toughness test samples, SFCM powder was ball milled overnight with a mixture of 0.6 wt% polyethylene glycol 600, 1.8 wt% ethylene glycol, and 0.6 wt% glycerol in isopropyl alcohol. It was then dried at 100 °C, ground by mortar and pestle, pressed uniaxially into rectangular bars at 30 MPa, then isostatically pressed at 30 MPa. Sintering followed by heating to 1340 °C for four hours at 3 °C/min with a one hour hold at 400 °C to allow the binders to burn out. This process achieved dense samples at 97% average theoretical density with no apparent flaws in the bars. Bars were then cut and sanded to final dimensions of 3 mm x 4 mm x 25 mm. The chevron notch was cut using the jig described by Jenkins, Chang and Okura following the ASTM procedure. [37, 38]

2.1.3 Test Coupons

The ASL and half-cell coupons used in the flexural testing were made using tape casting. NiO and GDC powders in a 6:4 weight ratio were mixed with ethanol, toluene, and fish oil. The mixture was ball milled on a rotary mill for 24 hours. Polyvinyl butyral and benzyl butyl phthalate were added, followed by another 24 hours milling. The resulting slurry was degassed and tape cast using a 700 micron blade height. The tapes were left to dry overnight before being cut into 12 cm by 12 cm squares. Three squares were stacked and hot pressed at 49 °C and 2 tons for 30 minutes. Following this lamination, the tape was cut into rectangular coupons measuring 25mm by 10mm. The coupons were fired at 1400 °C for 4 hours. Final bar dimensions were 8.16 by 24.15 by 2.98 mm on average. For half-cell coupons GDC electrolyte slurry was prepared and tape cast with a 40 micron blade height. After the lamination of the three ASL layers, a single layer of electrolyte tape was laminated to the ASL using the same pressure for 2 hours. The sintering procedure for half-cell coupons was identical to the ASL coupons. Samples were polished with 600 grit sandpaper on edges before testing.

Tape casting was used to create test coupons of porous SFCM-GDC ASL and SFCM-GDC/GDC half-cells. Using ethanol as a solvent, SFCM-GDC was ball milled with polyvinyl butyral, benzyl butyl phthalate, 12 µm poly(methyl methacrylate) (16 wt% with respect to SFCM-GDC), and Menhaden fish oil. The tape was cast to a thickness of 110 µm on Mylar then laminated using a hot press to a final thickness of 660 µm. Dense GDC was casted to 30 µm and laminated to the top of

the SFCM-GDC to create half-cells. Individual coupons were cut from the green tape, sintered at 1200 °C for four hours, with holds at low temperatures to burn out organic binder and pore former. The final thicknesses were measured to be 400 μm for the SFCM-GDC ASL and \sim 20 μm thick GDC electrolyte. Test coupons had their edges sanded to remove defects left from cutting, following ASTM standards. [39] Top and bottom surfaces were not sanded to preserve possible defects left from tape casting procedure, which would be representative of industrially manufactured SOFCs.

2.2 X-Ray Diffraction

X-ray diffraction (XRD) was used confirm phase purity of the SFCM after synthesis and during the testing process. A Bruker D8 Advance with LynxEye was used with a Cu K α source. A step size of 0.02° was used with a dwell of 0.8 s was used. Rietveld refinement was performed on samples as synthesized, after oxidation in pure O₂ and after reduction in pure H₂ at 600 °C for both. GSAS-II was used to perform the refinement calculations and VESTA was used to visualize the unit cell. [40, 41]

2.3 Density Measurements

Density of test bars was determined using the Archimedes method following Equation 2.1, where ρ is the density of the sample, ρ_{liq} is the density of the liquid used, w_{dry} is the weight of the sample before submersion and w_{sub} is the apparent

weight after submersion. Isopropyl alcohol was used as the liquid for measurements and samples were placed in a vacuum chamber while submerged to ensure all open porosity was filled before measuring the submerged weight.

$$\rho = \frac{w_{dry}}{w_{dry} - w_{sub}} \rho_{liq} \quad (2.1)$$

2.4 Thermogravimetric Analysis

Mass of samples during oxidation and reduction cycling was measured using a Cahn D200 microbalance. The samples were placed in a crucible suspended from a platinum wire attached to the microbalance and enclosed by an alumina tube inside a furnace. Heating control was achieved with a PID loop and temperature measurement done by a K-type thermocouple placed immediately below the sample inside the alumina tube. Gas flow was controlled at a constant 50 sccm by mass flow controllers. 21% O₂ in dry N₂ was used for the oxidizing condition while 3% H₂ in N₂ humidified with 3% H₂O was used for the reducing conditions.

To understand the degree of reduction in test coupons, thermogravimetric analysis (TGA) was used to measure the mass loss as a function of time, temperature, and gas environment. A Cahn D200 microbalance was used to measure the weight changes as the sample was heated in a furnace with controlled atmosphere. A small section of test coupon was placed in a crucible suspended from the balance and heated to 650 °C at 10 °C min⁻¹, the environment was switched from 50 sccm of 21% O₂ in N₂ to 50 sccm of 3% H₂, 3% H₂O in N₂. Mass, temperature, and p_{O₂} measurements were taken at 30 second intervals. p_{O₂} measurements were taken using

a calibrated YSZ sensor placed immediately before the sample. Samples were cooled at $10\text{ }^{\circ}\text{C min}^{-1}$ in the desired atmosphere.

Changes in mass of SFCM were measured by a Cahn D200 microbalance with the sample suspended down a quartz tube into a furnace. The furnace used to heat the sample and quartz tube was controlled by a PID controller with a K-type thermocouple placed immediately below the sample inside the quartz tube. Gas flow was controlled at consistent 50 sccm by mass flow controllers, which mixed dry nitrogen, oxygen, hydrogen, and humidified nitrogen to obtain the p_{O_2} desired. Measurements of the p_{O_2} were taken by a calibrated YSZ oxygen sensor at $800\text{ }^{\circ}\text{C}$ located before the sample. Again, intermediate p_{O_2} ranges were not able to be tested due to SFCM's incompatibility with the species required to create that environment.

Further details of the operation of the TGA and its various components are given in Appendix ??.

To prepare the sample, it was pre-weighed, wrapped in platinum wire and suspended from the balance, placed in the furnace with simulated air (21% O_2 , 79% N_2) flowing. Once the mass had stabilized, the furnace was heated to $800\text{ }^{\circ}\text{C}$ to allow for any organic contamination to burn off. After the mass stabilized at this elevated temperature the sample was introduced to various environments. The mass of the sample would be noted only after steady state had been reached for a condition. After testing a sample, a bar of alumina was cut to the same dimensions as the sample and the process was repeated to obtain a blank which could be subtracted from the measurements to remove buoyancy effects.

Oxygen non-stoichiometry was calculated using Equation 2.2, where $\Delta\delta$ is

the change in oxygen stoichiometry, MW_{SFCM} is the molecular weight of SFCM ($208.74 \text{ g mol}^{-1}$), MW_O is the molecular weight of oxygen (16.0 g mol^{-1}), w_{sample} is the weight of the sample, and Δw is the weight change as recorded by the TGA. The oxygen vacancy concentration ($[\text{V}_\text{O}^{\bullet\bullet}]$) is calculated using Equation 2.3, where ρ is the density of SFCM and N_A is Avogadro's number. To calculate the oxygen vacancy concentration, the non-stoichiometry of SFCM at a point needs to be established. For this work, based on the plateau present in the data at oxidizing conditions, it is assumed that in a pure oxygen environment ($\log(p_{\text{O}_2}) = 0$) all oxygen vacancies are filled with no oxygen interstitial or surface species, thus $\delta = 0$.

$$\Delta\delta = \frac{MW_{SFCM}}{MW_O w_{sample}} \Delta w \quad (2.2)$$

$$[\text{V}_\text{O}^{\bullet\bullet}] = \frac{\delta\rho N_A}{MW_{SFCM}} \quad (2.3)$$

2.4.1 Temperature Programmed Desorption

The effluent from the TGA was used as the inlet to a mass spectrometer (MS) to perform temperature programmed desorption. The sample was prepared as before and heat treated to remove any carbon contaminants but was allowed to cool under simulated air. It was then heated to 800°C at 5°C min^{-1} under a 50 sccm flow of nitrogen as the MS measured the 32 m/z signal which corresponded to O_2 desorption. Additional m/z signals were monitored to observed for other species.

2.5 Conductivity

Electric conductivity across p_{O_2} was measured using the four-wire technique and a Stanford SR 830 lock-in amplifier. A bar shape sample was used with dimensions of $6.46\text{ mm} \times 3.3\text{ mm} \times 1.3\text{ mm}$. Gold paste was used as a current collector, and the current range was between 0.005 to 0.05 A. A yttria-stabilized zirconia (YSZ) oxygen sensor operating at $800\text{ }^\circ\text{C}$ was used to monitor the changes in oxygen partial pressures. Intermediate p_{O_2} ranges were not tested due to an incompatibility between SFCM and the CO and CO_2 required to obtain those p_{O_2} .

Conductivity during redox cycling was measured using rectangular bars of SFCM and SFCM-GDC (2:1) composite. The samples were connected to a Keithley 2400 source meter by silver paste and wire. Using an in-house built reactor, the sample could be heated and the gas environment could be controlled. An initial measurement was taken after heating and 50 hours of exposure to 10% H_2 in N_2 , then the sample exposed cycled between air and reducing conditions over a period of 14 days.

2.6 Mechanical Testing

Measurements of mechanical properties were collected using a Tinius Olsen 10ST Universal Testing Machine (UTM) equipped with a 250 N load cell. Experiments where any samples would be tested at elevated temperatures or under reducing environments were conducted using a custom built three point flexural test fixture

placed inside a gas-tight chamber and furnace. Otherwise, a fully-articulating four point flexural test fixture was used. Samples tested at ambient conditions were placed on the appropriate testing fixture and loaded until failure. For samples tested at elevated temperatures the sample was heated in the test chamber at $10\text{ }^{\circ}\text{C min}^{-1}$, allowed to equilibrate for 20 minutes, then tested. Samples to be reduced were placed in the chamber, heated and exposed to reducing gas for 18 hours before being tested.

All tests conducted in the UTM (Tinius Olsen 10ST with 250 N load cell) were done at a rate of 0.2 mm min^{-1} with a 20 mm lower span. After measuring the force at fracture, stress was calculated using Equation 2.4 for 3-point flexural of rectangular samples, where σ_f is the stress at failure, F is the load at failure, L is the span of the fixture, b is the width of the sample and d is the thickness of the sample. At room temperature, the coupons or bars were loaded into the fixture and tested in batches of 5 per condition. Occasionally samples would break or be damaged before testing, reducing the sample set. For sample sets at elevated temperatures in ambient atmosphere, the samples were loaded into the front of the chamber, acting as a staging area, prior to heating the on fixture in the center. Upon transferring the next coupon from the staging area to the fixture, a 20 minute waiting period was used to ensure that the coupon had reached thermal equilibrium prior to testing.

$$\sigma_f = \frac{3FL}{2bd^2} \quad (2.4)$$

Following the completion of the set, all coupon pieces were cooled at a rate of $10\text{ }^{\circ}\text{C/min}$. For testing in reducing atmosphere, each coupon was reduced and tested individually. Half-cell coupons were tested in two orientations, “electrolyte-up” and

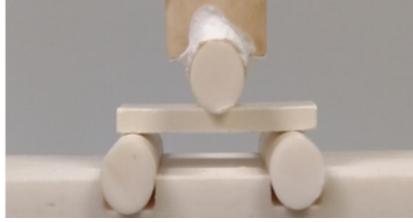


Figure 2.1: Assembled Alumina 3-Point Bend Fixture

“electrolyte-down.” These orientations cause the dense electrolyte layer to experience compressive or tensile stress. The “electrolyte-up” orientation subjects the electrolyte to compressive stress and vice-versa.

2.6.1 Development of Mechanical Test Apparatus

To develop a testing apparatus capable of simulating the various conditions experienced by operating SOFCs, appropriate materials were chosen based on thermal and chemical stability criteria. Alumina was chosen to build the bend fixture due to its chemical stability and high hardness. The sample rests crossways on two stationary 6.35 mm diameter rods which are placed in troughs separated by 20 mm. The upper half of the fixture consists of a 6.35 mm rod used to apply stress to the sample from the UTM crosshead. Both pieces are attached using silica-based cement to 300 mm rods attached at the anchor points of the UTM. Assembly and alignment is assisted with the use of a 3D printed jig to ensure repeatability. The complete alumina, 3-point bend fixture is seen in Figure 2.1 with the two bottom rollers, a sample, and the top roller making contact, applying flexural stresses to the sample.

To create the atmosphere control system, a combination of standard and custom vacuum system parts were used to enclose the bend fixture. For the main body

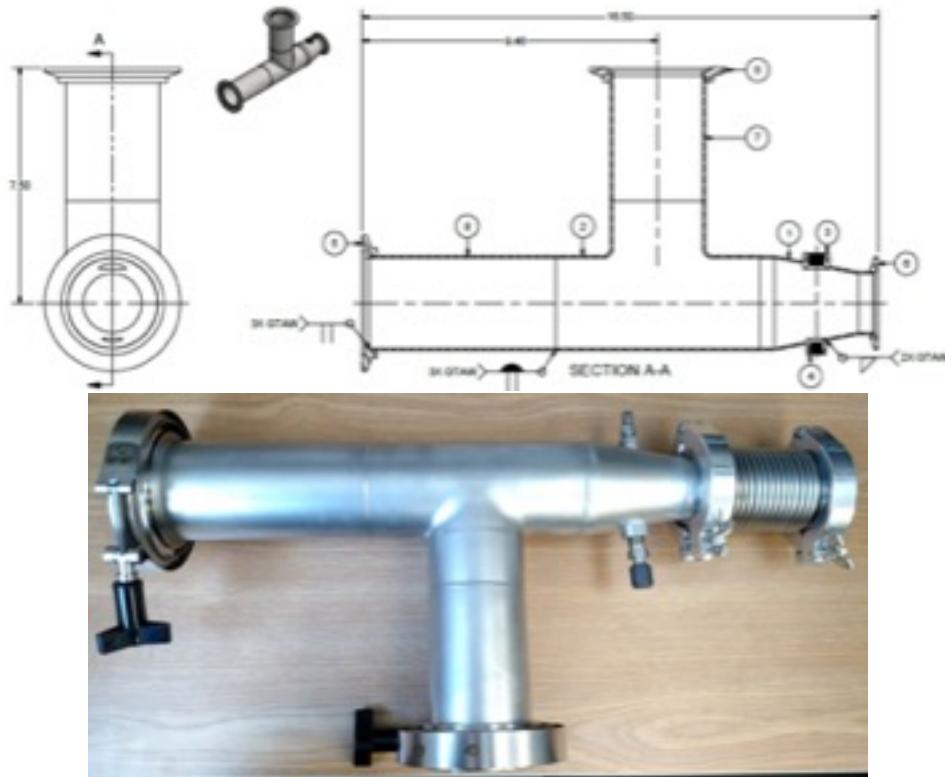


Figure 2.2: Design drawing and photo of mechanical testing atmospheric chamber

of the chamber, a 3-inch inner diameter, stainless steel tee was made with QF80 and QF50 flanges. Welded NPT ports on top and bottom allow for gas inlet and outlet and the introduction of a thermocouple through a Swagelok Ultra-Torr fitting. Figure 2.2 shows the engineering drawings and photo of the custom pieces from A&N Corp. Flexible bellows allow for the motion of the cross-bar to be translated into the fixture, requiring the subtraction of the spring force to be removed during analysis. The alumina fixture was then inserted through each end and the chamber incorporated into the furnace with QF50 to 1/2 inch tube adapters that fit over the alumina rods.

To heat the fixture and chamber, a 1 ft. cube was assembled from steel plates with cutouts on top, bottom and front. Steel wire mesh was used to create a framework inside to hold nickel-chromium alloy heating elements. Silica-based wool insulation was packed between the center cavity and the case walls. A K-type thermocouple was inserted into the chamber and used with an external PID controller to cycle the heating elements. The furnace was placed on a supporting scaffold to hold it in position during tests. The test fixture, atmosphere chamber, and furnace were affixed together as one unit and inserted into the UTM. The design was compared and qualified to a regular steel 3-point bend apparatus to ensure accurate results at room temperature.

2.6.2 Atmospheric Treatment

To test individual coupons under reducing environments, samples were loaded into the test chamber at 400 °C, after previous pre-heating in the staging area. After 20 minutes, the temperature would be increased to the desired set-point of 650 °C at a ramp rate of $10\text{ }^{\circ}\text{C min}^{-1}$ and the gas would be switched first to N₂ to flush the chamber, then to humidified argon containing 3% H₂. It would then be allowed to sit for 18 hours before testing to allow reduction. After testing, the chamber would be flushed again with N₂ and cooled to 400 °C before opening to change samples and repeat the process.

If samples had been previously reduced as a batch by exposure to hydrogen in a tube furnace and cooled under hydrogen, the atmospheric treatment was shortened,

as supported by TGA results. Pre-reduced samples were loaded into the chamber at or below 400 °C, the chamber flushed N₂, then with H₂ in Ar. After this the chamber would be heated to 650 °C, the sample tested after a 30 minute period, and cooled back down to 400 °C. At this point, the chamber would be flushed with N₂. While the chamber maintained a positive pressure of N₂, it could be opened, a new sample loaded, closed, and allowed to flush. This procedure allows for the batch reduction of many test coupons and replaced the 18 hour reduction time with a shorter 30 minute period.

2.6.3 Fracture Strength

A loading rate of 0.2 mm/min was used for all strength measurements. Strength was calculated from the maximum force measured before failure according to Equation 2.5 or 2.4 depending on if the four point or three point fixture is used respectively, where σ is the strength, P is the maximum force, L is the span width of the test fixture, b is the width of the sample and d is the thickness of the sample. Equation 2.5 is for a fixture where the top span is 1/2 the width of the bottom span. [39]

$$\sigma = \frac{3PL}{4bd^2} \quad (2.5)$$

2.6.4 Fracture Toughness

The fracture toughness of chevron notched samples were measured using a loading rate of 0.001 mm/min. Fracture toughness was calculated from the maximum force using Equation 2.6 or 2.7 for four point or three point fixtures. [38, 42] Y_{min}^* is

the shape factor as calculated by Equation 2.8, S_o and S_i are the outer and inner spans, B is the width of the sample, W is the height of the sample, a_0 is the distance from the tip of the chevron notch to the bottom of the sample, and a_1 is the average distance from the side of the chevron notch to the bottom of the sample. Each sample was measured after failure, but in this study the approximate values were $S_o = 40mm$, $S_i = 20mm$, $B = 3.0mm$, $W = 4.0mm$, $a_0 = 0.80mm$, $a_1 = 3.8mm$. Fracture toughness was measured only under air at room temperature and up to 600 °C. Under reduction the fracture toughness bars of SFCM would spontaneously fracture, preventing measurements under that condition.

$$K_{Ivb} = Y_{min}^* \left[\frac{P[S_o - S_i]}{BW^{3/2}} \right] 10^{-6} \quad (2.6)$$

$$K_{Ivb} = Y_{min}^* \left[\frac{P}{BW^{1/2}} \right] 10^{-6} \quad (2.7)$$

$$Y_{min}^* = \frac{0.38742 - 3.0919(a_0/W) + 4.2017(a_1/W) - 2.3127(a_1/W)^2 + 0.6379(a_1/W)^3}{1.0000 - 2.9686(a_0/W) + 3.5056(a_0/W)^2 - 2.1374(a_0/W)^3 + 0.0130(a_1/W)} \quad (2.8)$$

2.7 Scanning Electron Microscopy (SEM)

After fracture, SEM images were taken of the fracture surfaces. This was to observe the nature of the crack as it propagated, determining if it was transgranular or intergranular fracture, to find any flaws on the fracture which would lead to failure, and to observe pore geometry. Images were captured on an Hitachi SU-70 Schottky field emission SEM. Fracture surface images were mounted vertically and GDC samples were sputter coated with gold. Observations of porosity were taken using epoxy infiltrated samples and polished using down to a 0.25 µm diamond solution.

Chapter 3: High Temperature Mechanical Behavior of Porous Ceria and Ceria-Based SOFCs

3.1 Introduction

For planar SOFCs, a standard stack design consists of a number of square cells separated by metal interconnects. [43] These interconnects act as current collectors, gas channels, and separators to prevent fuel and air from mixing. Sealing of these stacks is accomplished by compressing the layered structure which includes sealing material. As more compressive force is applied to the stack, the quality of the seals improves. However, any cell which is not perfectly flat will experience flexural stress as a result and this can lead to cells fracturing. Due to the brittle nature of the ceramic materials that constitute SOFCs, a very thorough understanding of the mechanical limits of these devices is critical to their successful deployment.

The vulnerability of SOFCs to mechanical failure is a well-known issue. However, much of the research into this phenomenon has focused on yttrium stabilized zirconia (YSZ) based devices. [21,44,45] This material has been the standard for the field and has desirable mechanical properties but requires high temperatures to function well. As efforts are made to lower SOFC operating temperatures, a shift to ceria-based

electrolytes has occurred. Less attention has been given to the mechanical behavior of doped ceria materials across SOFC operating conditions.

Much of the study of fracture in ceramic materials has been done on technical ceramics for medical applications and for coating metal components. [46–53] These materials are optimized for fracture toughness and durability and very rarely experience temperatures above a few hundred degrees Celsius. The fracture surface analysis and correlations between microstructure and strength described for these materials are a valuable starting point for investigating fuel cell materials. However, there is a lack of extensive investigation into the properties of IT-SOFC materials at their expected operating temperatures and environments.

Efforts have been made to comprehensively examine the mechanical properties of SOFC materials and the effect of non-standard conditions on those properties. Nakajo et al. conducted a wide ranging study of materials used in anode supported SOFCs which included some attention to temperature and atmosphere effects. [45] While providing a solid base of material knowledge, this work did not fully cover materials beyond YSZ and there remains a need for further testing, especially for doped ceria. Flexural strength and Young's modulus measurements for gadolinium doped ceria (GDC) have been carried out in ambient conditions by Yasuda et al. [54] They characterized the effects of sintering temperature on density and the resulting mechanical properties. Further testing of this material system at elevated temperatures and anode gas reducing environments must be done to fully understand the mechanical behavior of GDC.

This study presents the results of a broad range of flexural tests involving the

materials used in ceria-based anode supported SOFCs along with the assembled half-cells. Using a purpose-built temperature controlled environmental chamber installed in a universal testing machine (UTM), porous doped ceria bars, anode support layers (ASL) composed of nickel and doped ceria cermet, and half-cells composed of an ASL and a doped ceria electrolyte were tested. The various test conditions used included expected operating temperatures (450 °C–650 °C), and both reducing and oxygenated atmospheres. These variations in test conditions are important because these cells must maintain their integrity from when they are first placed in a sealed stack to when they reach operating conditions. In particular, this study was intended to determine at what point in their life SOFCs are most vulnerable to mechanical failure and the mechanisms involved. Additionally, the effect of the anode-electrolyte interface on flexural strength was explored.

3.2 Results

3.2.1 Porosity and Pore Former Choice

As expected, GDC bars with greater porosity displayed lower flexural strength values as compared to less porous samples. The relationship between porosity and strength followed an exponential trend. This behavior is well established for porous ceramics and is described by Equation 3.1, where σ_f is the flexural strength with porosity, σ_o is the flexural strength without porosity, η is a geometric constant dependent on the system, and P is the volume percent porosity of the sample. [55]

$$\sigma_f = \sigma_o e^{-\eta P} \quad (3.1)$$

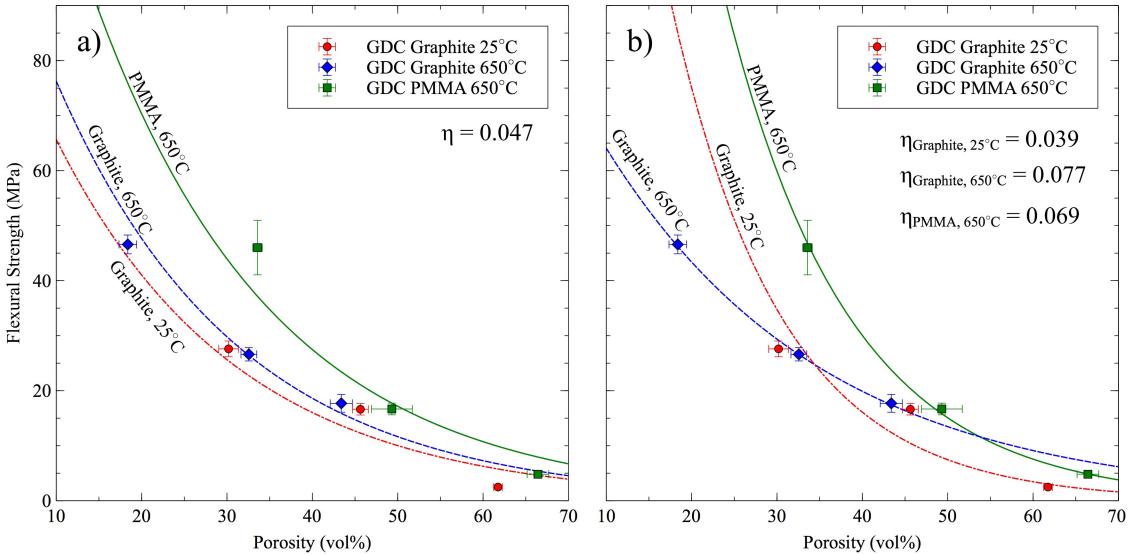


Figure 3.1: Flexural strength-porosity dependence for porous GDC10 at 650 °C and 25 °C using spherical PMMA or graphite flake pore former: a) Measured strengths fitted using fixed geometric constant; b) Measured strengths fitted using different geometric constants

The data was fit using this equation and both a fixed geometric constant, η , of 0.047 (Figure 3.1a) and a variable geometric constant (Figure 3.1b). Chi-squared values, indicating the divergence of the fit from the measured data for each sample set, were calculated for the two fitting methods and are displayed in Table 3.1.

The deviance from the fit curve for the fixed η method is quite large relative to the variable η fit for both sample sets tested at 650 °C. Both fitting methods are fairly poor in the case of the room temperature graphite set. The poor fitting results in this case is due to the >60% porosity samples. The flatter pore geometry caused by the graphite likely leads to comparatively greater strength loss at high porosity due to likely greater pore connectivity. The comparison between the high

Table 3.1: Chi-squared values for fixed η and variable η fits of porous GDC strength data (Figure 3.1

Sample Set	χ^2	χ^2
	(Fixed η)	(Variable η)
GDC PMMA 650 °C	3.406	0.062
GDC Graphite 650 °C	0.668	0.008
GDC Graphite 25 °C	3.575	5.063

temperature data sets is the most useful, given this, and the fact that SOFCs operate at this temperature. The poor fit of fixed η models suggests that a material-specific η value may be an oversimplification of the porosity-stress relationship in ceramics.

Samples tested at higher temperatures displayed a slightly higher flexural strength. Generally, it is expected that increasing temperature lowers strength of dense ceramics once plastic deformation processes become activated at high temperatures. [56] Below the temperature at which plastic deformation starts to occur, temperature has a minimal effect on flexural strength of dense ceramics. For dense GDC it has been demonstrated that the strength decreases by 19% between room temperature and 800 °C. [13] Porosity changes this behavior, by essentially creating a composite material, keeping the rate of change constant or increasing with temperature. [57] For porous GDC, thermal expansion places a compressive stress on any surface flaws that are sites for crack initiation and propagation. Additionally, the porosity helps reduce bulk stresses caused by thermal expansion. These two

effects increase the material's fracture strength more than any decrease from plastic deformation at 625 °C.

Cross-sectional SEM of the fracture surfaces of the porous GDC samples, both PMMA and graphite, is shown in Figure 3.2. Figure 3.2a shows individual spherical pores left by the PMMA while Figure 3.2b has an inter-connected network of long pores left by the graphite flakes.

The differences in quality of fit between Figure 3.1a and Figure 3.1b shows that the geometric factor is microstructurally dependent. Samples with porosity formed using PMMA spheres showed significantly greater strength as compared to samples with graphite-formed porosity. This can be explained by the effect of pore geometry on crack initiation and propagation in the ceramic. If a crack enters a pore, the pore can now be considered the new crack tip. The energy required to advance the crack is highly dependent on the geometry of the tip. For a spherical pore, this geometric factor is maximized and results in higher resilience to fracture. This is summarized by Equation 3.2 where σ_o is the stress at the crack tip, c is the length of the crack tip and ρ is the radius of curvature of the crack tip. [58]

$$\sigma_f = \frac{\sigma_o}{2} \sqrt{\frac{\rho}{c}} \quad (3.2)$$

Based on the mechanical behavior of these samples, porous ceramics should be designed and constructed such that the pore geometry is as low aspect ratio as possible to maximize strength.

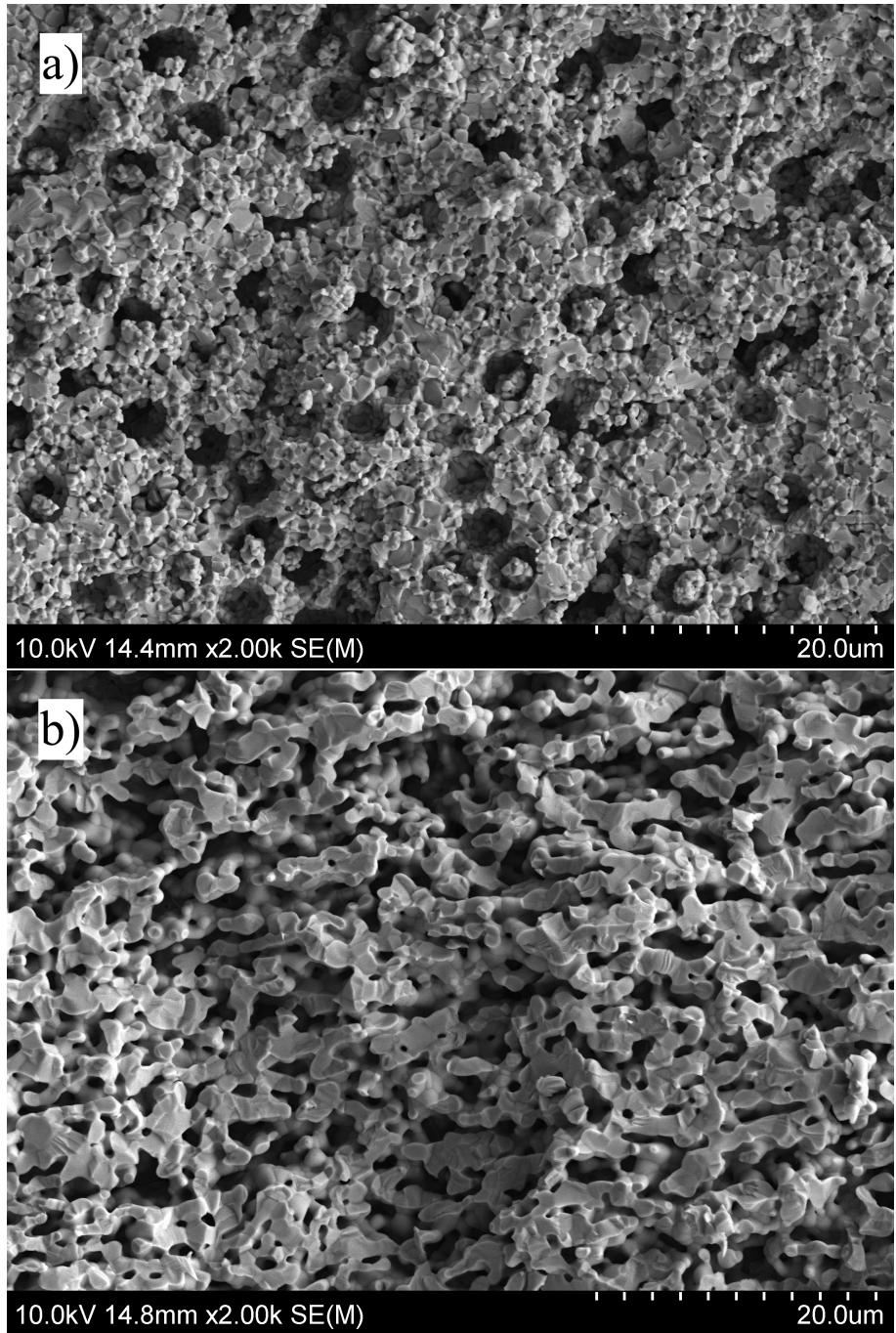


Figure 3.2: SEM micrographs of porous GDC bars made using: a) PMMA pore former; b) graphite flake pore former

3.2.2 Flexural Stress Orientation

Figure 3.3 shows the measure flexural strengths of a number of SOFC coupon samples. Anode support coupons were tested along with half-cells consisting of anode supports and electrolyte layers at temperatures ranging from 25 °C to 650 °C. The highest temperature test was repeated with coupons under reducing conditions (3% H₂, balance Ar).

All unreduced sample types showed increased strength at elevated temperatures with little difference between types at a given temperature. The high-temperature reduced coupons displayed large differences in strength depending on the orientation of the sample. Tests in which the dense electrolyte layer was placed in compression resulted in the highest strength values, while the samples were weakest when the electrolyte was placed in tension. In reduced samples, the anode support layer becomes a ceramic-metal composite and is therefore somewhat elastic while the electrolyte remains a brittle ceramic. The electrolyte-in-compression condition maximizes the mechanical performance of the coupon by placing the layers in their preferred stress state.

Figure 3.4 shows the box plots for the unreduced SOFC coupon samples. A Student's t-test was used to statically determine if the means of samples tested at different conditions were equal based on 95% confidence. There is no discernible difference in the strength between the three sample types at each temperature. At elevated temperature there was a significant difference in the modulus of ASL only samples and samples with electrolyte in tension with a p-value of 0.0091. This

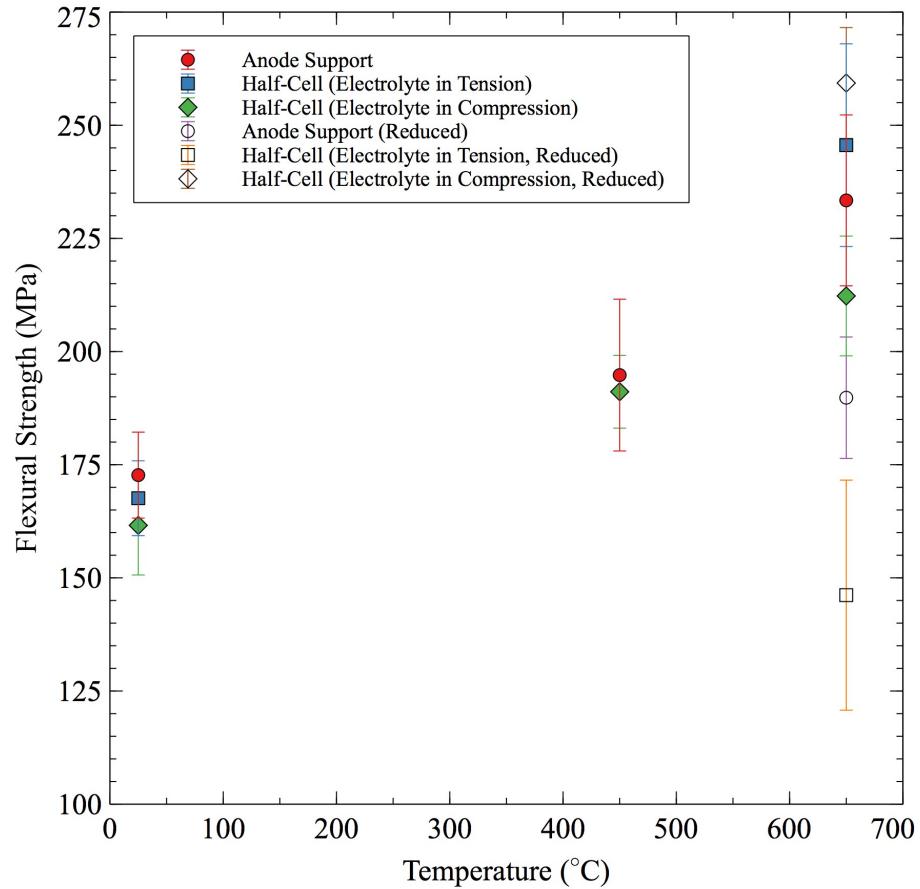


Figure 3.3: Temperature dependent strength of Ni-GDC anode supports and half-cells in both air (filled data points), and reducing atmosphere (3% H₂, balance Ar, hollow data points), tested in both “electrolyte-up” and “electrolyte-down” orientations resulting in the electrolyte layer experiencing tension and compression, respectively

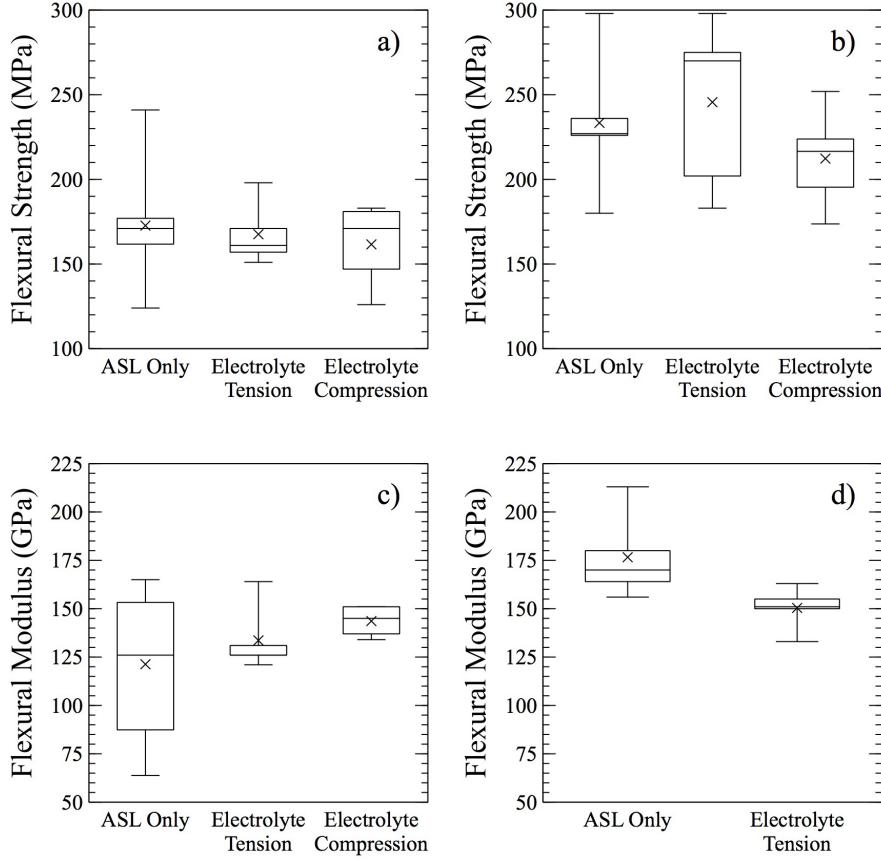


Figure 3.4: Flexural test measurements of coupons sample sets: a) Strength at 25 °C b) Strength at 650 °C c) Modulus at 25 °C d) Modulus at 650 °C

difference was not present at room temperature. As the materials reach elevated temperature, small differences in elasticity become magnified due to different thermal effects on dense and porous layers.

Additionally, SEM analysis of the fractured half-cells showed very good adhesion between layers (Figure 3.5). Delamination is a common failure mode in layered ceramics and one that would be particularly damaging to SOFCs due to resulting ionic conductivity loss between layers. [59] In the half-cell coupons, it was clear that

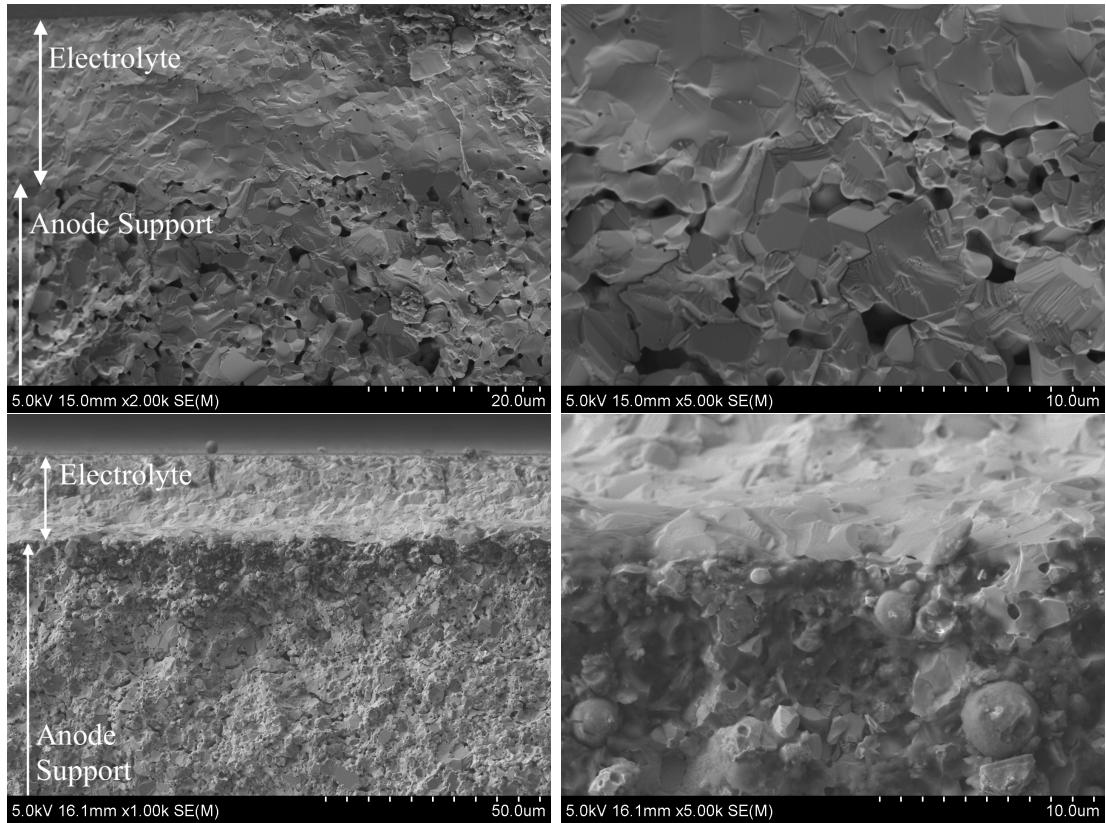


Figure 3.5: SEM micrographs of unreduced and reduced half-cell fracture

surface: tested in air at 25 °C (top) and tested in reducing atmosphere at 650 °C (bottom), showing good adhesion between anode and electrolyte

the fracture plane contained mixed transgranular and intergranular fracture. Some grains were sheared through while others remained whole. The striations visible in Figure 3.5 are characteristic of fracture proceeding through a grain, while other grains remained whole.

3.2.3 Reduction and Strength

Mass loss of SOFC half-cell coupons exposed to reducing atmosphere at various elevated temperatures is shown in Figure 3.6. Mass loss is attributed to the reduction

Table 3.2: Summary of fit parameters for reduction of NiO-GDC/GDC half-cell coupons under 3% H₂, 3% H₂O, 94% N₂ at different temperatures.

Temperature (°C)	Exponential Rate (h ⁻¹)	Asymptote (Mass %)
550	0.0463	89.78
575	0.135	89.09
600	0.182	88.36
650	0.512	87.75
750	0.821	87.87

of NiO, used as a precursor in fabrication, to Ni metal, which serves as the catalyst for fuel oxidation and electronic conductor, and the reduction of ceria to CeO_{2-δ}. Ni-GDC/GDC half cells showed the expected trend of increasing reduction rate at higher temperatures. Reduction curves were fit to an exponential decay and the summary of parameters is shown in Table 3.2. At 650 °C and 700 °C the reduction occurs very quickly, reaching steady state values after 18 hours. 550 °C showed a much slower mass loss than even 575 °C. At temperatures lower than 550 °C the kinetics are slow enough to tolerate a brief exposure to oxygen without re-oxidizing the sample. Of note in the reduction is that each temperature appears to approach a different asymptote, showing that the amount of NiO and GDC reduced at steady state is dependent on the temperature of the cell. This will affect the mechanical properties of the cells as it changes both the porosity and amount of nickel metal in the samples.

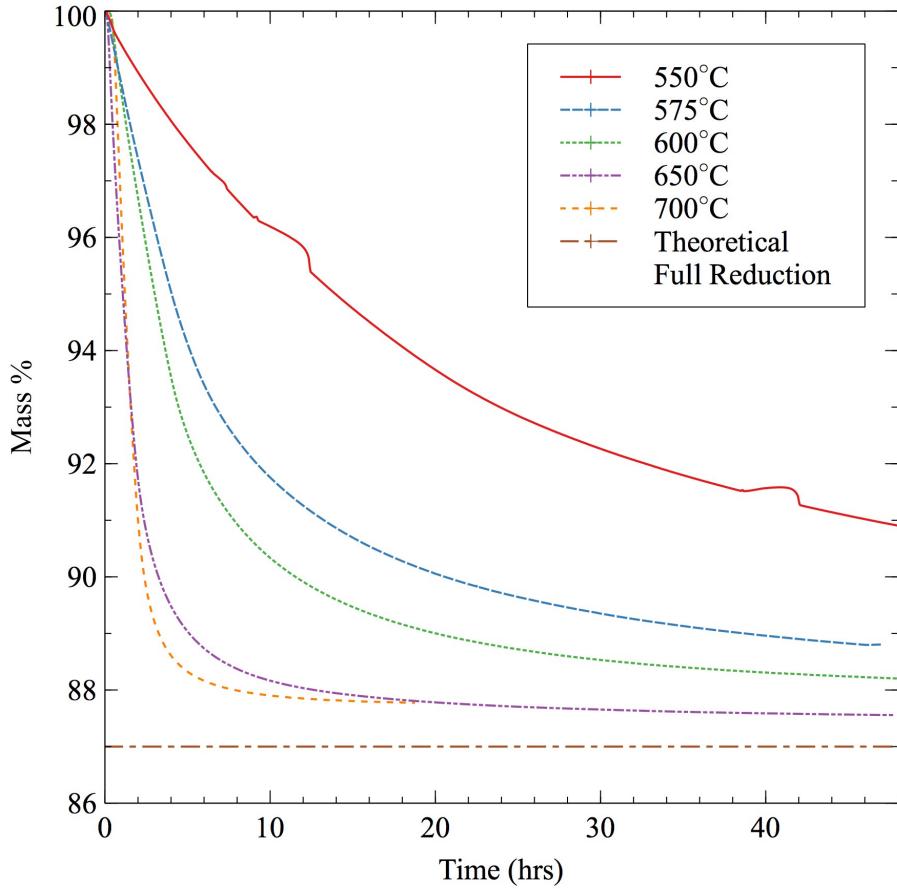


Figure 3.6: Thermogravimetric analysis curves for Ni-GDC/GDC half-cells showing mass loss over time at temperatures ranging from 550 °C to 750 °C in 3% H₂ 3% H₂O balance N₂, 50 sccm flow.

Ni-GDC/GDC fuel cells are sensitive to re-oxidation once reduced. Re-oxidation causes fracture of cells due to the large volume difference between NiO and Ni. [45] Based on the results from the TGA, the reaction rates below 300 °C are sufficiently slow to allow for the exposure of a reduced cell to oxygen without detrimental results. [60] To confirm this a sample was measured by TGA to ensure no mass gain during oxygen exposure and the strength of samples were tested to ensure there was no discernible difference between reduction methods.

For TGA analysis, a piece of a half cell was heated to 650 °C with a $10\text{ }^{\circ}\text{C min}^{-1}$ ramp rate and reduced. It was then cooled while still under reducing atmospheres. Once below 100 °C, it was exposed to simulated air (21% O₂) for 18 hours, placed back into reducing atmosphere, and heated back up to 650 °C. Figure 3.7 shows the change in mass overlaid onto the temperature and oxygen partial pressure experienced by the sample. Following this treatment, the cell showed no mass gain during the oxygen exposure and continued to reduce at the same rate as before once it was returned to the initial conditions. This shows that cells which are cooled appropriately do not re-oxidize and could be handled in between a batch reduction of cells, and their assembly into a stack configuration.

Box plots of flexural strength and modulus of the in-situ reduced and batch reduced coupons are shown in Figure 3.8. Mechanical strength of half-cell coupons which had been reduced in-situ with an 18 hour reduction time showed no difference in strength when compared with coupons which had been previously batch reduced, cooled, and reheated under reducing environments. Both of these sample types showed a dramatic decrease in strength and Young's modulus compared to the un-

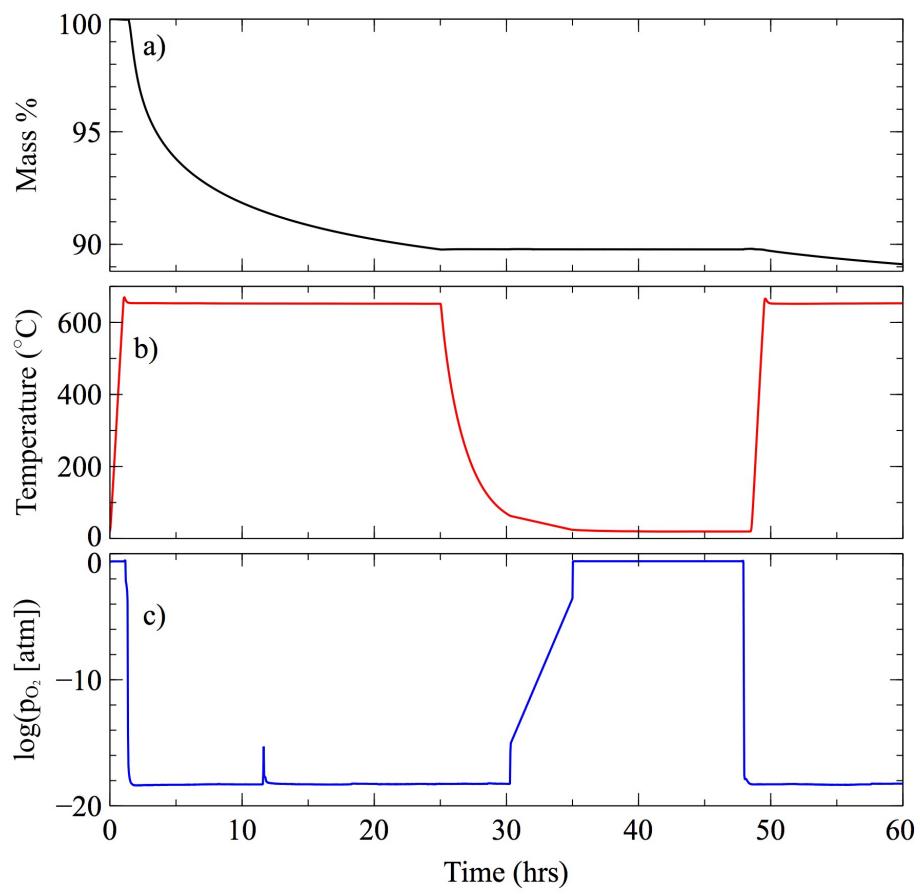


Figure 3.7: Thermogravimetric analysis of Ni-GDC/GDC half-cell showing no mass gain after reduction and exposure to room temperature and simulated air: a) Mass loss of interrupted reduction with exposure to ambient condition; b) Temperature of sample during cycle c) Oxygen partial pressure measured at the sample during cycle

reduced samples, but no statistically significant difference between treatments. The decrease in strength and modulus is due to the increase in porosity and conversion of NiO to metallic nickel in the cells. The decrease in variation between measurements in reduced cells is likely due to the increased number of large voids that form during NiO reduction.

The statistical similarity of the properties of the samples reduced via each method indicates that planar, Ni-GDC/GDC based SOFC ASL and half-cells are able to be safely reduced, cooled, and handled in ambient conditions without leading to damaging re-oxidation of the nickel anode material. This resilience could enable some degree of large-batch reduction of cell anodes prior to SOFC stack assembly, leading to more rapid startup and a greater degree of sealing control.

3.3 Conclusions

A temperature and atmosphere controlled three-point bend fixture was designed and built for use in a universal testing machine. SOFC coupons and component materials were evaluated for flexural strength at room temperature and IT-SOFC operating temperatures. In addition, the effects of porosity percent and pore geometry on flexural strength in GDC were investigated. Furthermore, the impact of temperature on the reduction rate of NiO in NiO-GDC SOFC anodes was examined along with the resilience to re-oxidation at ambient conditions of this SOFC component.

Pore geometry had a significant impact on the flexural strength of GDC10, with spherical pores showing the greatest resistance to fracture. This supports the

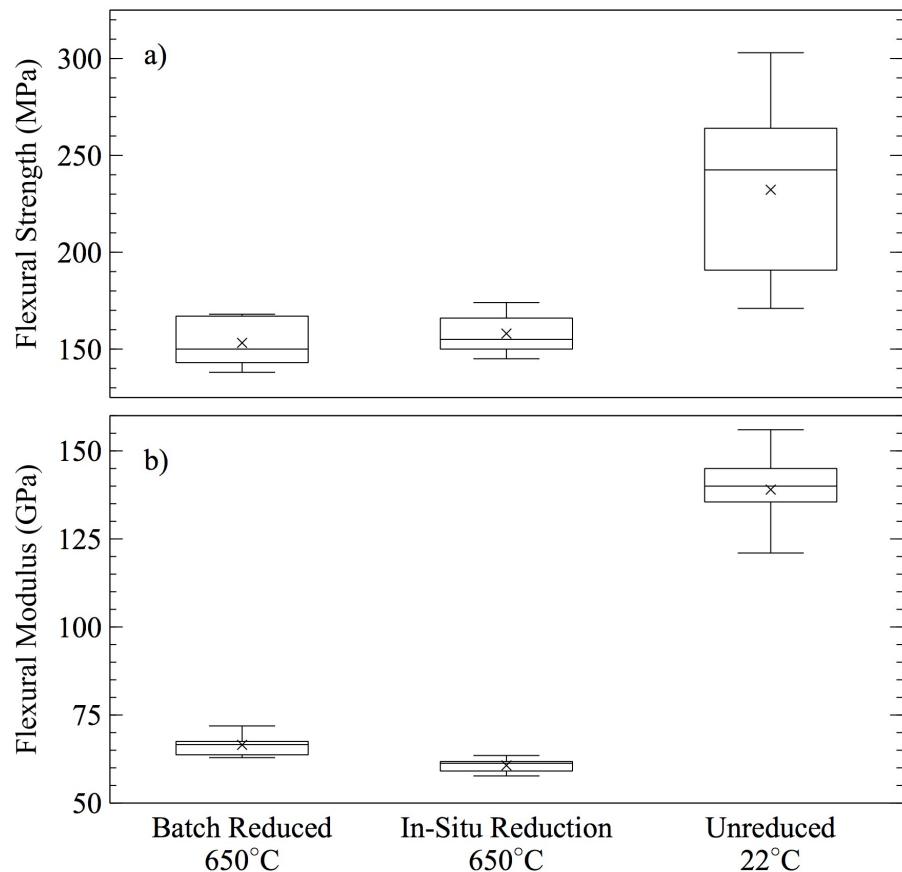


Figure 3.8: Flexural properties, a) strength and b) modulus, of Ni-GDC/GDC half-cell coupons after reduction via two different methods, compared to unreduced cells. Strength and modulus show decrease upon reduction but no significant difference between methods.

concept of pores acting as the new crack tip once a crack has advanced to the pore. Additionally, samples tested at 650 °C were stronger than those at room temperature. This is likely due to localized compressive stresses from thermal expansion of the material. This hypothesis is further supported by the results of testing NiO-GDC anode support coupons and half-cells. Coupons tested at 25 °C, 450 °C, and 650 °C displayed a linear strength dependence with temperature. There was no statistical difference in strength between anode support layers and half-cells composed of anode support and electrolyte at a given temperature in air. Half-cells in which NiO was reduced to Ni by exposure to H₂ at 650 °C displayed significant differences in strength when the electrolyte layer was subjected to compressive stress as opposed to tensile stress. Placing the ceramic electrolyte in compression and the metal-ceramic composite anode in tension resulted in the highest strength.

The reduction temperature of NiO-GDC/GDC half-cells was shown to have an effect on the rate of NiO reduction and amount of NiO reduced. At lower temperatures, the oxidation rate of Ni-GDC is slow enough that the anode can be exposed to air for significant periods below 100 °C. Any re-oxidation, combined with the cooling and re-heating of a cell back to 650 °C, showed no effect on the mechanical properties when compared to cells which had been reduced in-situ at 650 °C. These results indicate that it is possible to reduce and cool cell components to an extent without any additional effects to mechanical properties, allowing for more flexibility during cell manufacturing, stack assembly, and with quality control screenings.

This work leads to three important conclusions for the mechanical properties of

GDC-based SOFCs using Ni anodes. Porous GDC used in anode supported SOFCs should be fabricated such that the pore geometry is spherical as this maximizes energy required to advance a crack through the ceramic. Care should be taken in stack construction to ensure any out-of-plane cells are placed to compress the electrolyte and place the reduced anode in tension so as to lower the chance of fracture. Finally, it is possible to reduce the anodes of Ni-GDC SOFCs and then handle them at ambient conditions for quality control and stack assembly. This will remove a degree of variability from the manufacture of cells and stacks.

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